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THE TITANIUM-ALUMINUM-BERYLLIUM SYSTEM'S POTENTIAL FOR METAL MATRIX COMPOSITES

L. A. Jacobson
L. E. Tanner

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PREFACE

This effort was performed under the sponsorship of the Defense Advanced Research Projects Agency, in response to ARPA Order 5634. Dr. Phil Parrish is the DARPA Program Manager, and Dr. Gil London of the Naval Air Propulsion Test Center serves as the DARPA technical representative.

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INTRODUCTION

The history of development of metal matrix composites has been extensively concerned with the introduction of reinforcement phases by physical mixing techniques. Such approaches often lead to a number of undesirable composite characteristics, including non-uniform mixing and uncontrollable interfacial reactions between the constituents. The objective of the present work is to explore metallurgical techniques, including rapid solidification, as alternative ways of achieving a useful dispersion of reinforcing phases in a metal matrix. The Ti-Al-Be system appeared to have interesting possibilities for obtaining useful dispersions of intermetallic compound reinforcements with either aluminum or titanium as the matrix material.[1,2] This report presents the results of initial investigations into the phase relations and solidification microstructures of alloys in the Ti-Al-Be system. The main emphasis during the current reporting period has been on alloy compositions in the aluminum-rich corner of the ternary system.

EXPERIMENTAL

Fifteen alloy compositions with from 1% to 5% Ti and 4% to 15% Be were prepared by arc melting in the form of 20-gram buttons. Several melts of each button were performed to insure complete melting and to achieve the maximum homogeneity. Metallographic sections were prepared for each of the compositions, and portions of the button were reserved for rapid solidification processing. Each composition was quenched as a thin foil using an arc-hammer splat apparatus, and several splats of each were prepared. In order to test the thermal stability of the rapidly solidified alloys, Vickers hardness was measured, using a 25-gram load, for each splat quenched sample. Then, a portion of the splat was treated at 400°C for 100 minutes, and the hardness was measured following this treatment.

Optical metallography was used to examine the as-cast buttons, and portions of each button were homogenized at 600°C for several days. Homogenized samples were also examined by optical metallography in order to determine the amounts of each phase present for each composition.

Transmission electron microscopy was employed to examine the alloy microstructures representative of the types of treatments investigated so far.

RESULTS

Two alloy compositions containing 4% Ti with 8% and 10% Be had high hardness, in the vicinity of 300 VHN, and did not suffer a drastic drop in hardness after heat treatment at 400°C for 100 minutes. Post-heat treatment values ranged from 250 to 280 VHN. The hardness values of all the samples in the as-splat and heat-treated condition are given in Table 1.

Optical metallographic observations on the aluminum-rich alloys in the as-cast condition are summarized as follows. All but one of the compositions show evidence of a eutectic structure; to identify the specific components of the eutectic will require transmission electron microscopy (TEM). One of the compositions, 8%Be-2%Ti, shows what appears to be two types of eutectic, and this should be further studied (Fig. 1). Another of the compositions shows evidence of a peritectic reaction (Al-8%Be-4%Ti), suggesting that it was on the titanium side of the ternary trough (Fig. 2). However, the composition 6%Be-4%Ti showed eutectic, when it should have also shown a peritectic-type microstructure (Fig. 3). This composition was re-done since there was some question as to whether it was correctly formulated, but the new sample also had evidence of a eutectic microstructure. All the compositions show primary plates of $TiAl_3$, although in the 8%Be-1%Ti composition there are very few, suggesting that if this composition were homogenized, they would dissolve (Fig. 4). Apparent volume fractions of plates seem to be consistent with the amount of Ti present in the sample. All compositions show evidence of what appear to be beryllide particles but it is impossible to determine from morphology alone which of the possible beryllides formed (Figs. 5 and 6). Additional x-ray diffraction and TEM will be performed to identify the phases that are present.

The 10%Be-4%Ti sample was examined in the as-splat condition using TEM, and was found to contain a uniform distribution of fine (50- to 200-Å)

Table 1. Hardness of aluminum corner compositions, as-solidified and heat-treated.

Alloy composition (atom percent)			Hardness (VHN, 25-g load)	
Al	Be	Ti	As-solidified	400°C, 100 min
95	4	1	181 (206)[a]	73 (81)
93	6	1	230 (251)	91 (77)
91	8	1	247 (203)	79 (75)
94	4	2	212 (206)	139 (153)
92	6	2	237 (206)	126 (125)
90	8	2	270 (237)	178 (172)
88	10	2	237 (247)	140 (125)
90	6	4	302 (274)	174 (224)
88	8	4	262 (317)	245 (258)
86	10	4	292 (345)	272 (283)
83	12	5	311	306
80	15	5	370	230
77	18	5	370	227
73	20	7	327	302
70.5	22.5	7	322	287
68	25	7	317	283

[a] Values in parentheses are repeat tests using different splat samples.

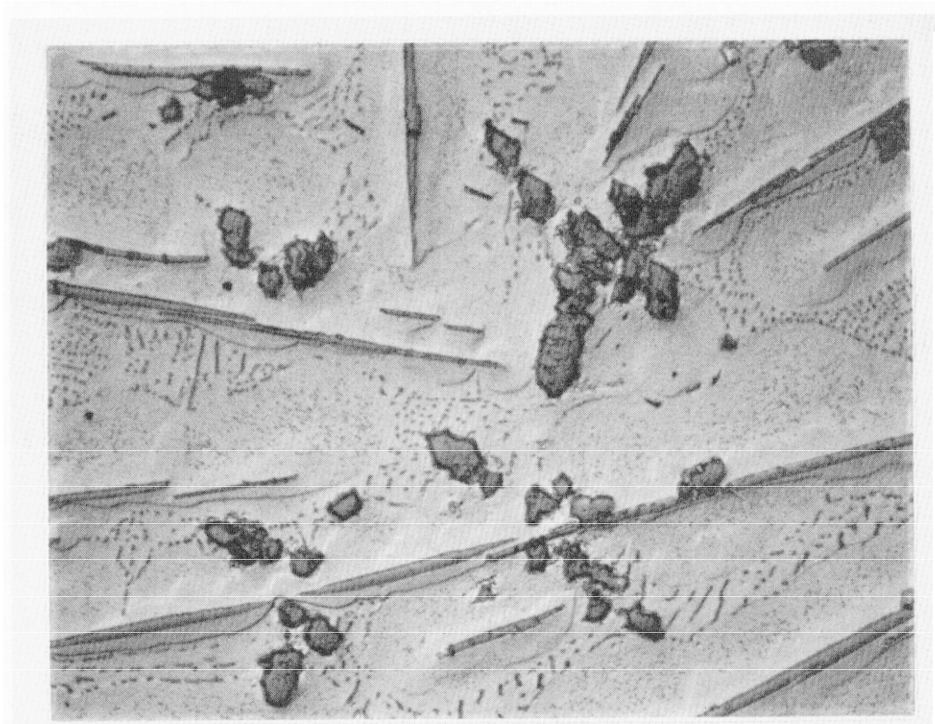

Fig. 1. 90%Al-8%Be-2%Ti as-cast sample showing primary beryllide (blocks) and $TiAl_3$ (plates) phases, along with two types of eutectic microstructure (500X).

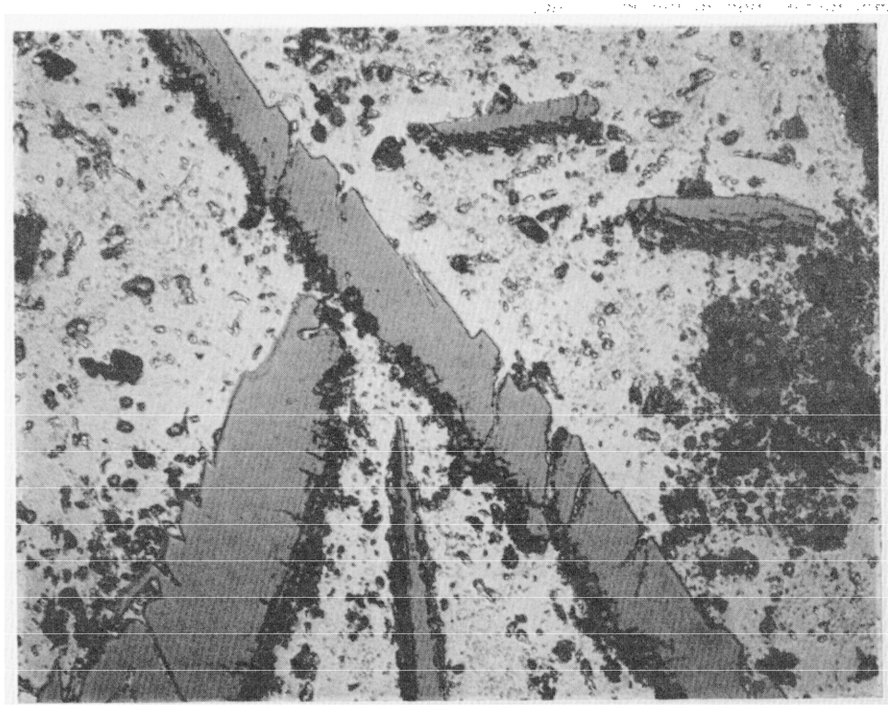

Fig. 2. 88%Al-8%Be-4%Ti as-cast sample with a peritectic-type microstructure (500X).

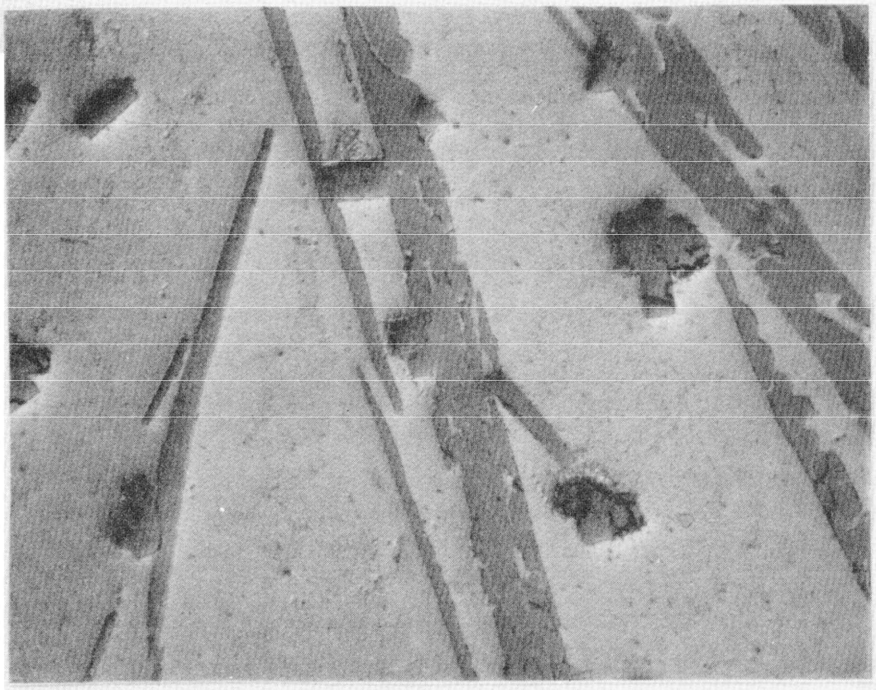

Fig. 3. 90%Al-6%Be-4%Ti as-cast, showing eutectic when peritectic was the expected microstructure type (500X).

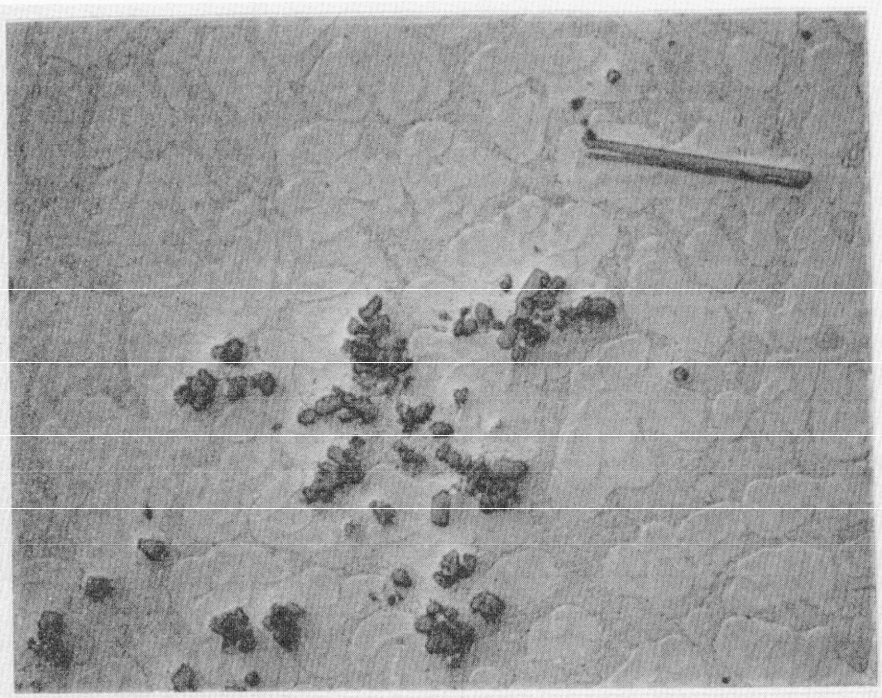

Fig. 4. 91%Al-8%Be-1%Ti as-cast, almost entirely free of $TiAl_3$ plates (500X).

Fig. 5. 78%Al-15%Be-7%Ti, expected to have peritectic structure but instead has a fine eutectic (500X).

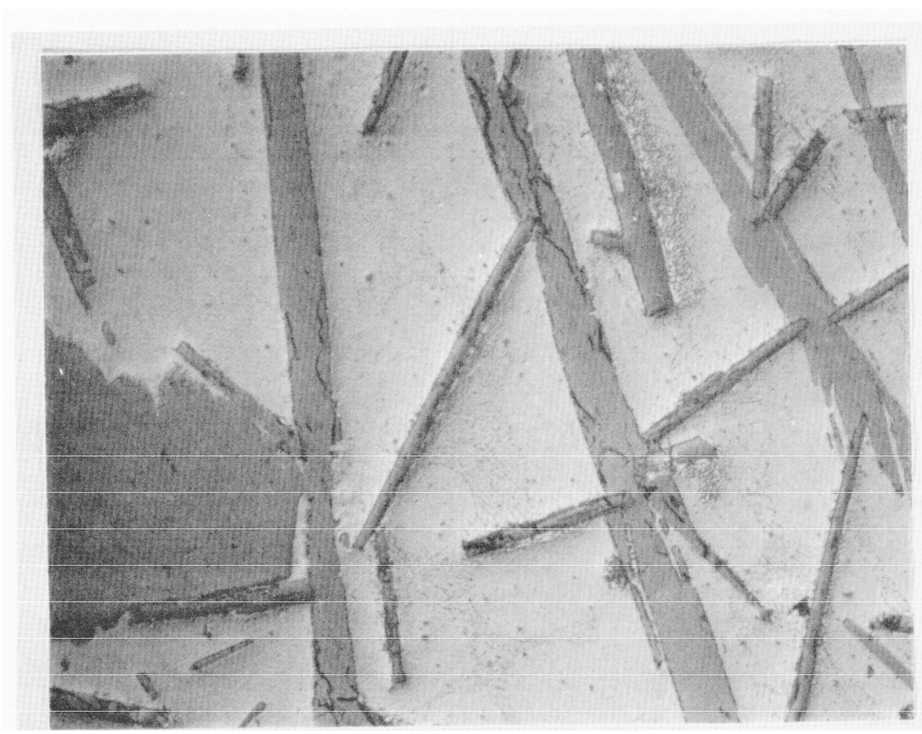

Fig. 6. 80%Al-17.5%Be-2.5%Al, intended to be in the two beryllide + Al phase field, but also has $TiAl_3$ plates (500X).

cubical particles of what appeared to be the $TiAl_3$ phase. An unusual aspect of these particles was that they were randomly oriented with respect to the aluminum grain in which they were contained. Another unusual finding was the lack of obvious particles of the other phase that was expected to be present, Ti_2Be_{17}. After annealing it was difficult to see any discrete particles due to a dense network of what may be interfacial dislocations. Still, only slight evidence for the presence of the beryllide could be found in the diffraction pattern. Figures 7a and 7b show the as-splat microstructure and the heat-treated microstructure of the 10-4 composition, along with their respective electron diffraction patterns. The random orientation of particles in the as-splat sample is obvious, both from the dark field view and the rings evident in the diffraction pattern. After heat treatment the structure is very difficult to interpret, with rings now not so evident in the diffraction pattern, and no features in the bright field micrograph that could be interpreted as enlarged cubical particles. We will be employing longer and shorter times at elevated temperature, and higher and lower temperatures in order to obtain better information on how the as-splat structure is changing.

DISCUSSION

Interpreting the solidification structure makes use of results obtained earlier on the Al-Be binary and on other alloy systems with a eutectic close to one of the constituents--Cu-Cu_2O, Fe-Mn-S, Ti-B, and Ti-C.[3] These results are not at present well accounted for by any theoretical model, but have the common feature of a uniform, fine distribution of particles, oriented randomly with respect to the matrix. It is felt that the lack of orientation may indicate that the second phase particles are forming just ahead of the moving solid interface, and are trapped by it as solidification proceeds. The conditions that are common to all the systems investigated by us include the aforementioned eutectic or near eutectic composition, low solubility of the added element or elements in the matrix phase (both equilibrium and metastable), and what must be a very low T_g such that an amorphous alloy does not form on rapid solidification. Conventionally solidified eutectics in such systems often show coupled growth and rod morphology, with clear orientation relationships between the rod phase and the matrix phase. At

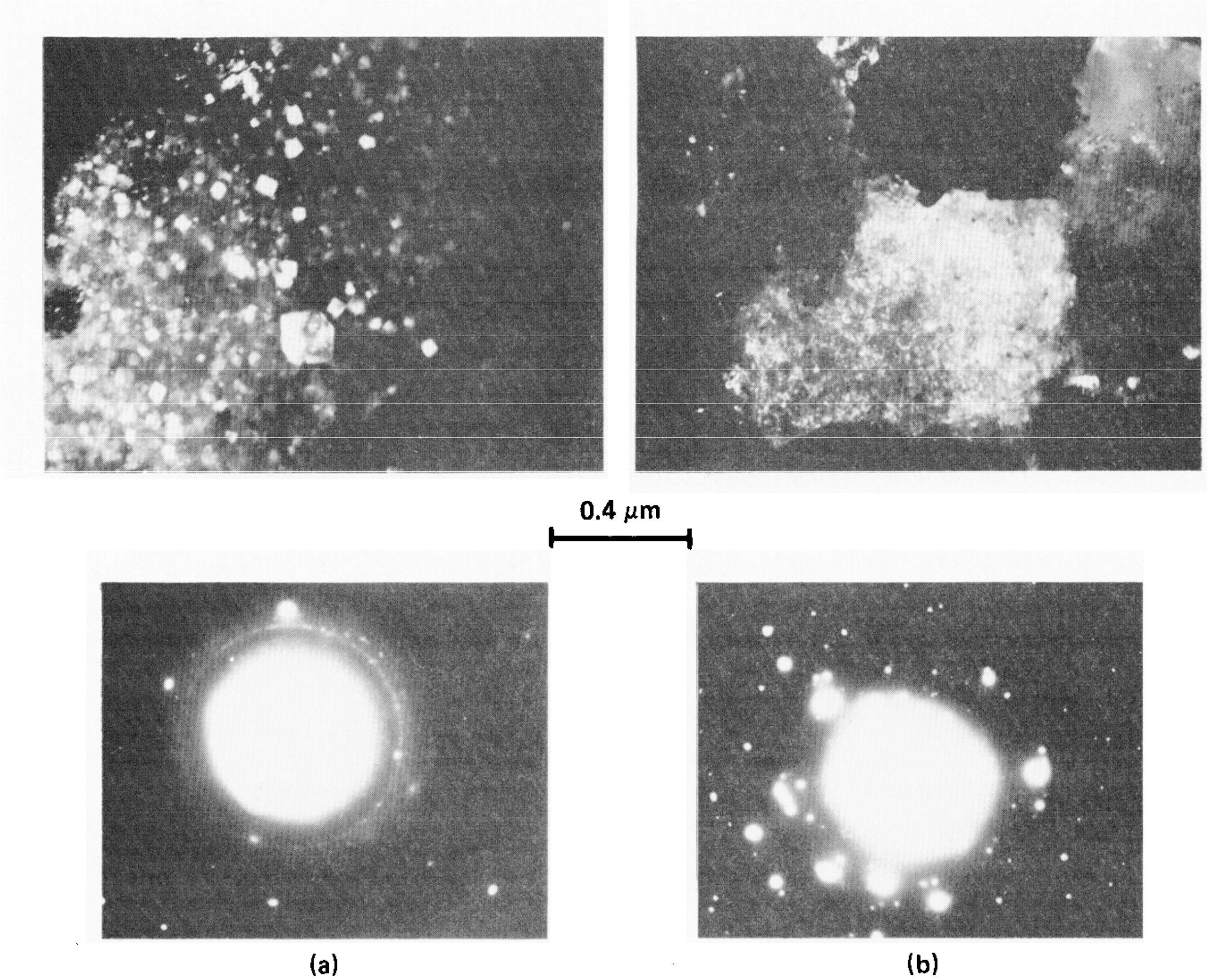

Fig. 7. 86%Al-10%Be-4%Ti transmission electron micrographs and selected area diffraction: (a) as-splat, dark field; (b) heat-treated at 400°C for 100 min, bright field.

intermediate solidification rates the rod structure breaks down into elongated ellipses, and finally spheres are formed, which also have an orientation relationship with the matrix. However, at some point, the particles no longer nucleate on the growing matrix surface, and their orientation is lost.

The 10-4 composition does not lie on a ternary eutectic point, but it does appear to lie on a ternary eutectic trough, which would explain the lack of primary phase particles in the splat microstructure. At such a trough, there would be no large temperature interval between the liquidus and the solidus temperature within which primary phases could form. It would then be possible for the liquid to be sufficiently undercooled for very rapid solidification to take place at roughly the initial composition.

The liquidus surface configuration in the vicinity of the aluminum corner is depicted in the perspective sketch shown in Fig. 8. What is not evident from this sketch is the detailed nature of the various transformations that could occur between the liquidus surface and the temperature below which all phases are solid. Figure 9 depicts an enlarged version of the aluminum corner of the ternary diagram, with the compositions of the alloys investigated so far, and the liquidus trough from the published ternary diagram plotted as a solid line. The broken line indicates the suggested location of the ternary eutectic trough based on our results, and the dashed lines are the probable boundaries between the various three-phase regions (Al-Be-$TiBe_{12}$, Al-$TiBe_{12}$-Ti_2Be_{17}, and Al-Ti_2Be_{17}-$TiAl_3$). All of the indicated alloys will be equilibrated at 600°C in order to check the positions of the phase boundaries. It will also be necessary to perform DTA or other thermal analysis in order to determine the sub-liquidus transformation temperatures for this portion of the phase diagram. Such information will help to interpret the rapidly solidified structures, and, more importantly, help to understand the progress of such structures toward equilibrium as they are exposed to elevated temperature.

We are continuing to explore the path of the eutectic trough, and have found that a 15-5 composition has an even higher as-splat hardness of 375 VHN. Compositions with 7% Ti show a lower as-splat hardness and are brittle, suggesting that the limit of potentially useful compositions has been exceeded, or that the eutectic trough does not proceed in the direction indicated by the existing ternary phase diagram.

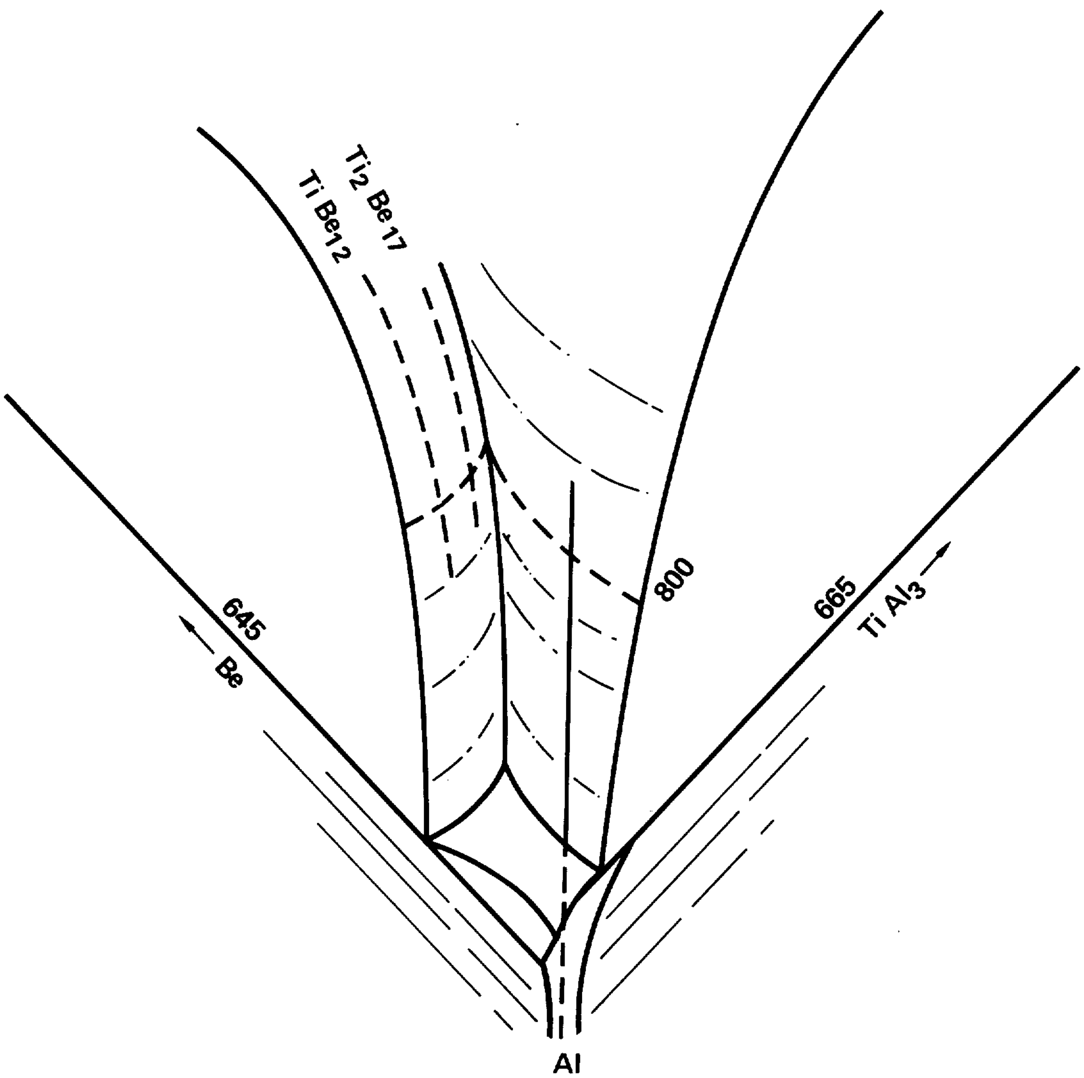

Fig. 8. Sketch of the liquidus surface perspective, aluminum corner.

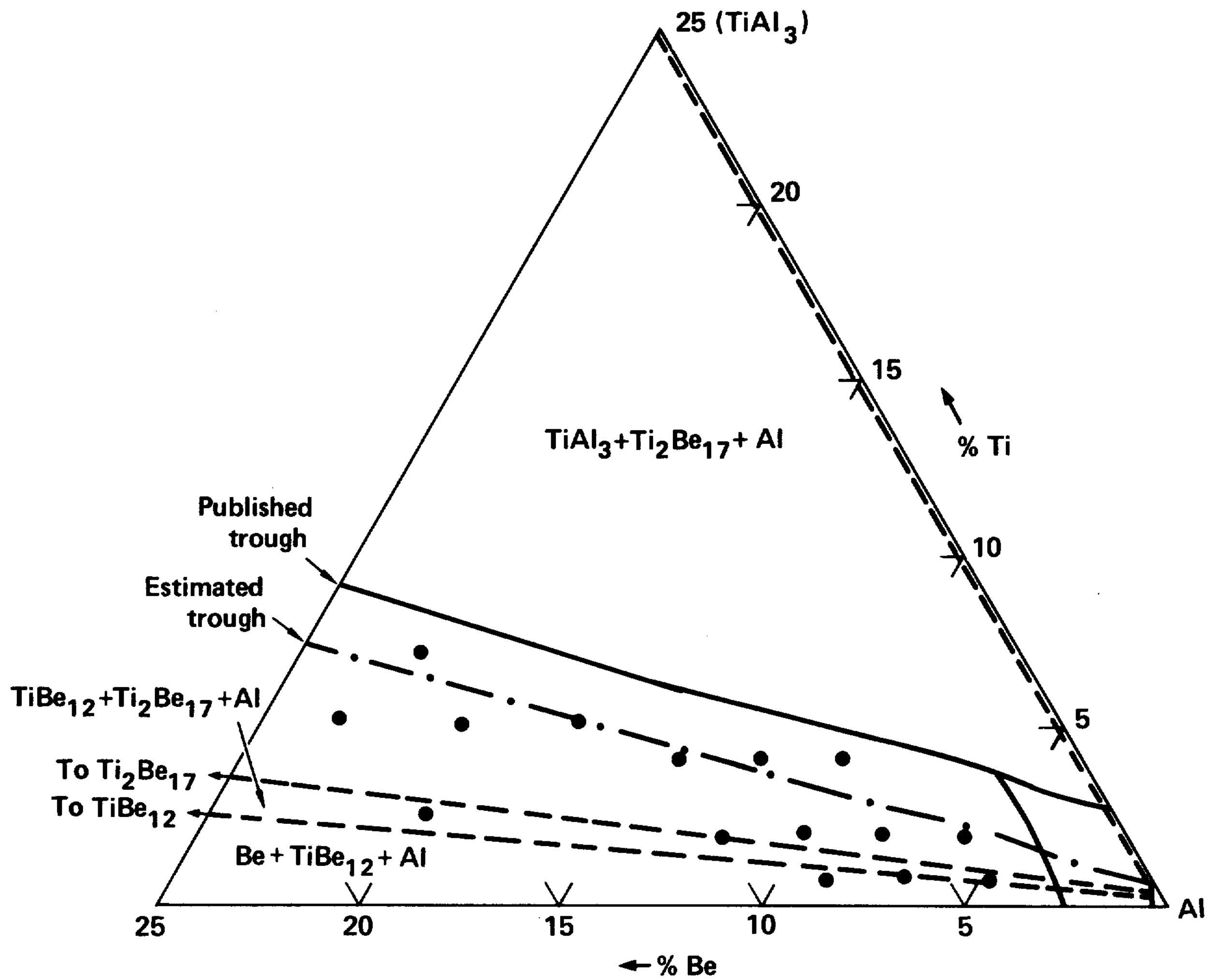

Fig. 9. Expanded aluminum corner showing principal features of the liquidus surface and the alloy compositions examined so far. Dashed lines are solid-phase field boundaries.

The randomly oriented $TiAl_3$ particles could be having a more significant effect on the strength of the alloy than would oriented particles. In the rapidly quenched binary alloys of Al and Ti, some metastable extension of solid solution is achieved, but it is not possible to prevent the peritectic formation of primary compound. Therefore, only small volume fractions of hardening particles can be realized and, since these are precipitated from solid solution, they are coherent with the matrix. Not only would the solid-state precipitation produce a much smaller volume fraction of particles, but perhaps the fact that they would be coherent with the matrix might not make them as effective strengthening agents as randomly oriented particles.

A number of titanium-rich alloys have also been made, and as of this writing we can only confirm that there are also some significant discrepancies between our findings and the published ternary phase diagram. These will be reported upon more thoroughly in the next progress report.

CONCLUSIONS

The initial period of this investigation has led to some interesting new findings regarding rapidly solidified microstructures, deviations from the published ternary phase diagram, and what appears to be excellent hardness retention after high-temperature exposure for several of the compositions in the aluminum-rich corner. More work is definitely required before any specific conclusions can be drawn concerning the equilibrium and the non-equilibrium phase relationships, microstructure formation and transitions, and possible relationships to potentially useful mechanical properties.

ACKNOWLEDGMENTS

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